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Some creep characteristics of beryllium-copper at 300degrees Celsius.

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SOME CREEP CHARACTERISTICS OF BERYLLIUM-COPPER AT 300°C

James G. Curry

BER	BERYLLIUM
COP	COPPER
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Letter on cover:

SOME CREEP CHARACTERISTICS OF
BERYLLIUM-COPPER AT 300°C

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This work is accepted as fulfilling
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ABSTRACT

The creep and creep-rupture characteristics of beryllium-copper have been examined. Hardness and metallographic data have been used to investigate structural changes occuring during tests on specimens in the solution-treated, aged and over-aged conditions for both cold worked and unworked sheet material. Variation of the condition of the alloy before testing has provided information on the effect of these changes on creep behavior under specific conditions of test.

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TABLE OF SYMBOLS AND ABBREVIATIONS

A	Solution-treated.
AT-316	Solution-treated and precipitation-hardened at 316°C. (600°F.) for three hours.
AT-600	Solution-treated and precipitation-hardened at 600°C. (1112°F.) for 70 hours, then slowly cooled at a rate of approximately 10°C./hr.
H	Solution-treated and cold rolled 37.1% (reduction in thickness).
HT-316	Same as H and precipitation-hardened at 316°C. (600°F.) for three hours.

INTRODUCTION

A study has been made of the creep and creep-rupture characteristics of a copper-beryllium alloy which is capable of pronounced age-hardening* on being subjected to relatively low elevated temperatures.

Beryllium-copper, as is well known, is hardened by the precipitation of CuBe from its solid solution in copper. Since Wilm's work in 1911, concerned with the first evidence of precipitation-hardening, a voluminous quantity of matter has been published on the mechanism of precipitation-hardening. Briefly stated, it has been proposed that hardening is caused by lattice strains created by lattice coherency between matrix and precipitate; much experimental evidence has been advanced to support this thesis(1).**

The following two characteristics of the Be-Cu alloys have been largely responsible for their spectacular development; (a) Beryllium-copper alloys are unsurpassed in their ability to withstand fatigue and wear and at the same time conduct electrical current under high temperature conditions. (b) They are unique among copper-base alloys in that they can

*The terms "age-hardening" or "precipitation-hardening" denote the simultaneous increase of hardness and other mechanical properties through heat treatment and the passage of time.

**The numbers in parentheses refer to the Bibliography at the end of this paper.

be worked in a relatively soft state and then brought to their final level of strength and hardness by a simple low temperature heat treatment.

Little previous information on the subject of creep in relation to age-hardening is available[2-5]. Jenkins, Bucknall, and Jenkinson[2], in their work on the creep-rupture characteristics of a copper alloy containing nickel and silicon, were probably one of the first to report such information.

The object of this investigation was to provide creep data on a precipitation-hardenable alloy of copper-beryllium, in various initial states and over a wide range of applied stresses, and at temperatures of 200°C., 300°C. and 400°C. The three temperatures selected would have provided data on the effect of structural changes due to various rates at which precipitation proceeded during reheating. However, due to experimental difficulties encountered at the outset of this investigation, it became necessary to limit the temperatures at which tests were conducted to one, namely 300°C. The difficulties were related to the design of the specimens employed and the extreme brittleness of the alloy, both of which will be discussed more thoroughly in a subsequent section of this paper.

EXPERIMENTAL PROCEDURE

1. TEST MATERIAL

Wrought beryllium-copper specimens were machined from 0.100-inch strip obtained from The Beryllium Corporation of Reading, Pennsylvania. The material, as received, was in part in the solution-treated condition, the balance being in the solution -treated and cold-rolled (37%) condition. The chemical composition and average room temperature properties of this alloy are presented in Tables I and II respectively.

TABLE I - CHEMICAL COMPOSITION*

Beryllium	2.00%	Silver	0.01%
Cobalt	0.22%	Tin	0.01%
Iron	0.14%	Nickel	0.01%
Silicon	0.12%	Chromium	0.01%
Aluminum	0.03%	Lead	0.00%
Zinc	0.03%	Copper	Balance

*Chemical analysis supplied by The Beryllium Corporation.

TABLE II - AVERAGE ROOM TEMPERATURE PROPERTIES#

Condition	Tensile Strength, 1000 psi	Yield Strength (0.01% offset), 1000 psi	Elongation, %	Rockwell Hardness Number	Mod. of Elastic. 10^6 psi
A	72	25	50	B 60	17
H	107	70	6	B 97.5	17
AT	175	80	5	C 41	19.5
HT	195	110	3	C 42.5	20

#Metals Handbook, 1948, ASM.

The five states chosen for testing were:

(a) solution-treated (as received).

(b) solution-treated and precipitation-hardened at 316°C. (600°F.) for three hours.

(c) solution-treated and cold rolled 37% (as received).

(d) solution-treated, cold rolled 37% and precipitation-hardened at 316°C. (600°F.) for three hours.

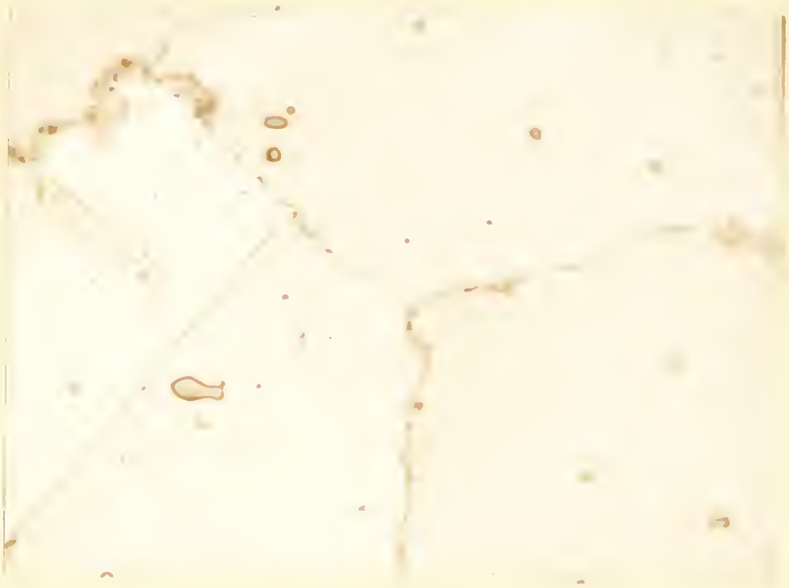
(e) solution-treated and overaged at 600°C. (1112°F.) for 70 hours and then slowly cooled at a rate of approximately 10°C./hr.

The following standard temper designations will be employed whenever reference is made to the above states: (a) A, (b) AT-316, (c) H, (d) HT-316, (e) AT-600. This notation is in accordance with the Table of Symbols and Abbreviations on page vi and is repeated here for convenience.

The appearance of the microstructures of these states before testing are shown in Figures 1 - 5, respectively. These micrographs show the presence of beta stringers which are due to incomplete homogenization. These beta stringers may have several undesirable characteristics; one of these is that the amount of beryllium in the alpha phase is reduced and is therefore not "available" for hardening. In addition, the "as-received" material (see Figures 1 and 3) showed a condition variously described as "grain boundary precipitate" or "grain boundary gamma". The cause or composition of this boundary phase is not fully understood and is the subject of considerable controversy[6]. However, this formation results, at least in part, from commercial mill practice. Its effect is to form clumps or masses of a phase along the grain boundaries that may have many undesirable consequences, among them excess distortion of the piece when heat treated. A slight amount of visible general precipitate within the grains and beryllites (beryllium-cobalt intermetallic compounds) were observed in the unaged material.

The low-temperature precipitation-hardening treatment, which produces the highest degree of strength and hardness, was accomplished in a salt bath because of its rapid and uniform heating characteristics. Following hardening, the specimens were quenched in cold water as this permitted immediate handling, minimized distortion and definitely fixed the end of the hardening time. In the aging of material which had not received any cold-working, considerable distortion was experienced in 70% of the specimens. This was probably due to the

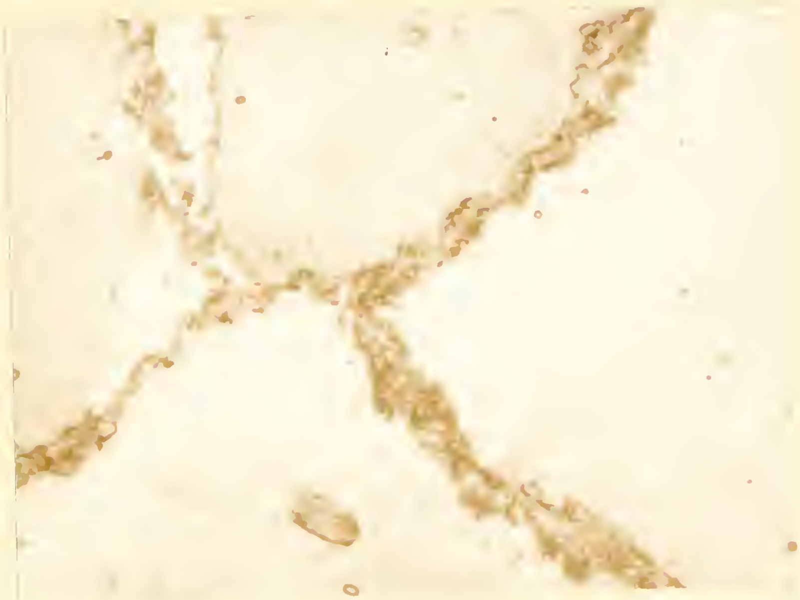
(a) X100



(b) X2000

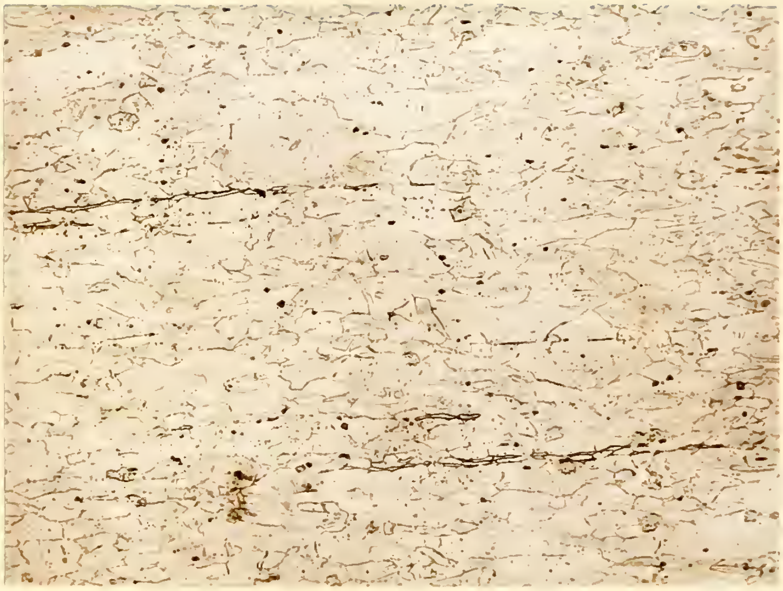
FIGURE 1--A Condition

(a) X100

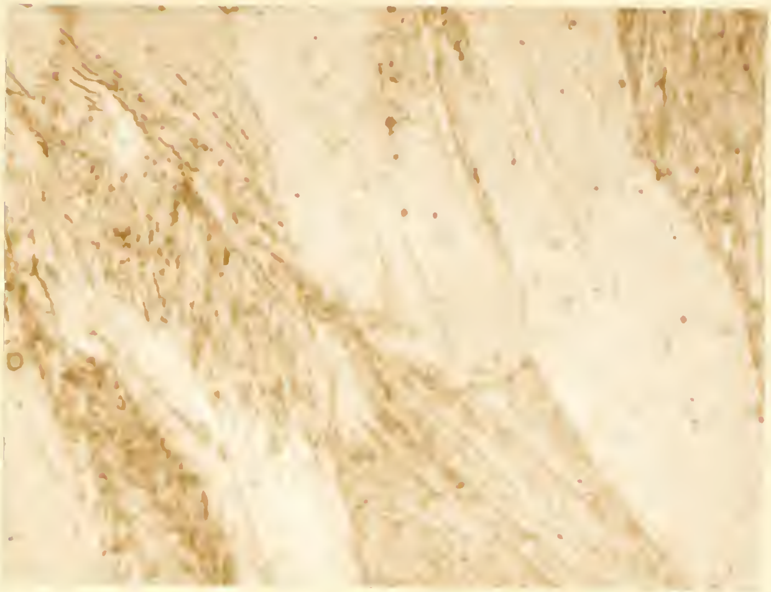


(b) X2000

FIGURE 2--AT-316 Condition



(a) X100



(b) X2000

FIGURE 3--H Condition

(a) X100

(b) X2000

FIGURE 4--HT-316 Condition

(a) X100

(b) X500

FIGURE 5--AT-600 Condition

grain boundary precipitate mentioned earlier. In the case of the cold-worked material negligible or no distortion was in evidence. The microstructures obtained by this treatment are shown in Figures 2 and 4.

The high-temperature over-aging treatment, which produces the lowest static strength and most ductile condition, was accomplished in a vertical tube furnace utilizing a helium atmosphere in order to prevent oxidation during the 70 hour treatment at 600°C. (1112°F.). Upon completion of this heat treatment, the temperature was lowered at approximately 10°C./hr. The resulting material (see Figure 5) was much less subject to structural change on further reheating than it was in the solution-treated or fully aged state in either the stressed or unstressed condition, precipitation of the second phase being much more complete.

2. Test Equipment

The constant-load creep tests reported in Table V were carried out on two conventional overhead lever-arm creep testing machines with 20:1 and 10:1 lever ratios. The lever arm had a counterweight feature which permitted the accurate taring of fixtures and the extensometer system. An overall view of one of the creep testing units is shown in Figure 6.

The measurement of strain along a 1-inch gage length was accomplished through the use of a mechanical extensometer system as shown in Figure 7. The dial gages mounted on the extensometers were capable of being read directly to 0.001 in., and an estimate to 0.0001 in. was possible. Since the details of the extensometer have been reported already they will not be reproduced here(7). Uniaxial loading was ensured by the insertion of a universal joint in the loading system in line with the specimen.

A manually operated screw jack was placed under the weight pan which supported the load. The load was applied by gradually lowering the jack which thus permitted the loading weights to hang free, resulting in a uniform application of load without impacts. The loading time was less than 30 seconds in all tests. All weights used were weighed on a fan type scale having a minimum division of 0.01 lb. with a range of 0 to 25 lbs. The dimensions of the original cross section of specimens were measured with a micrometer to 0.0001 in.

Conventional tube-type nichrome wound furnaces were used on the creep units. Taps located at close intervals along the

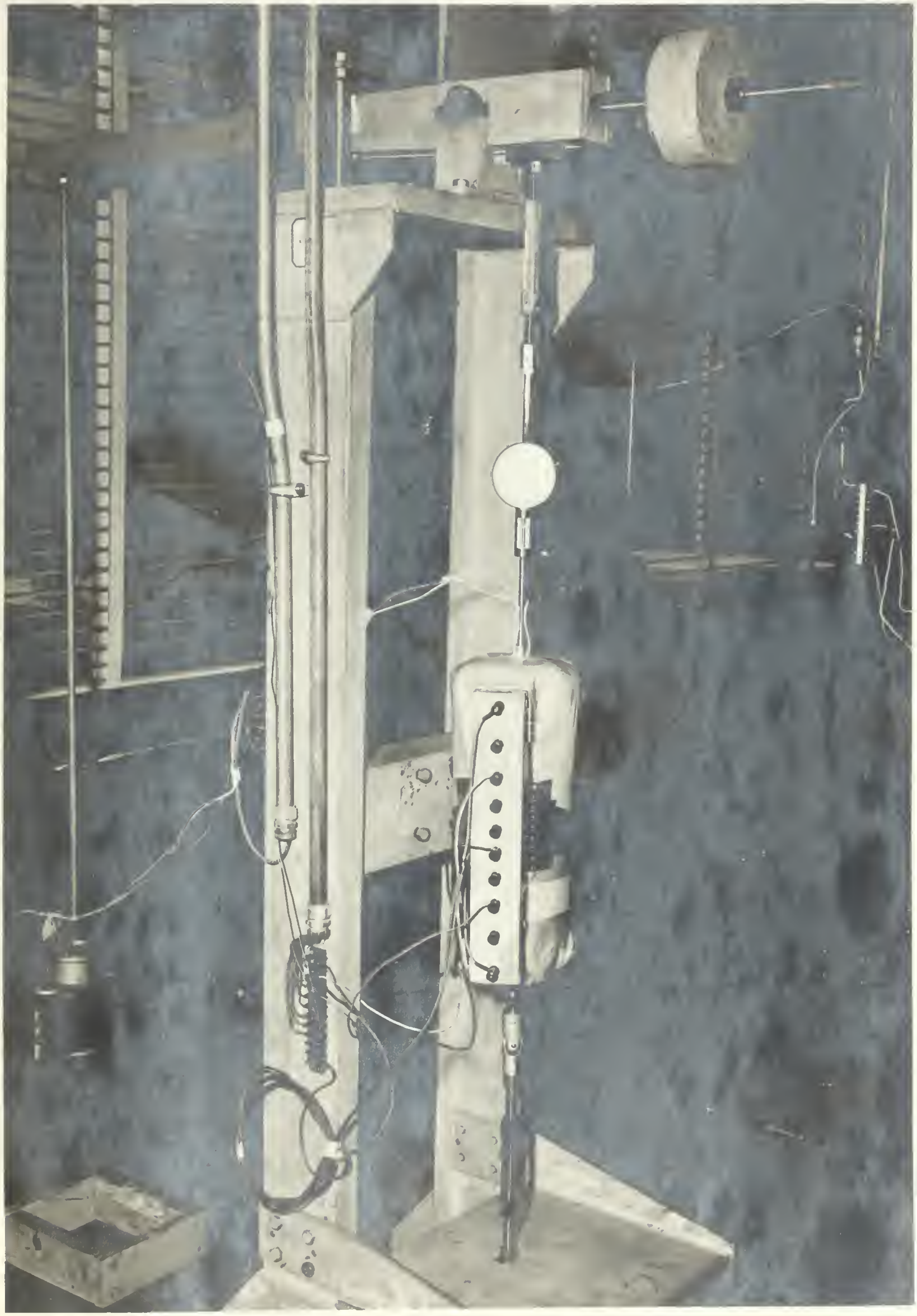


Figure 6 CREEP TESTING UNIT

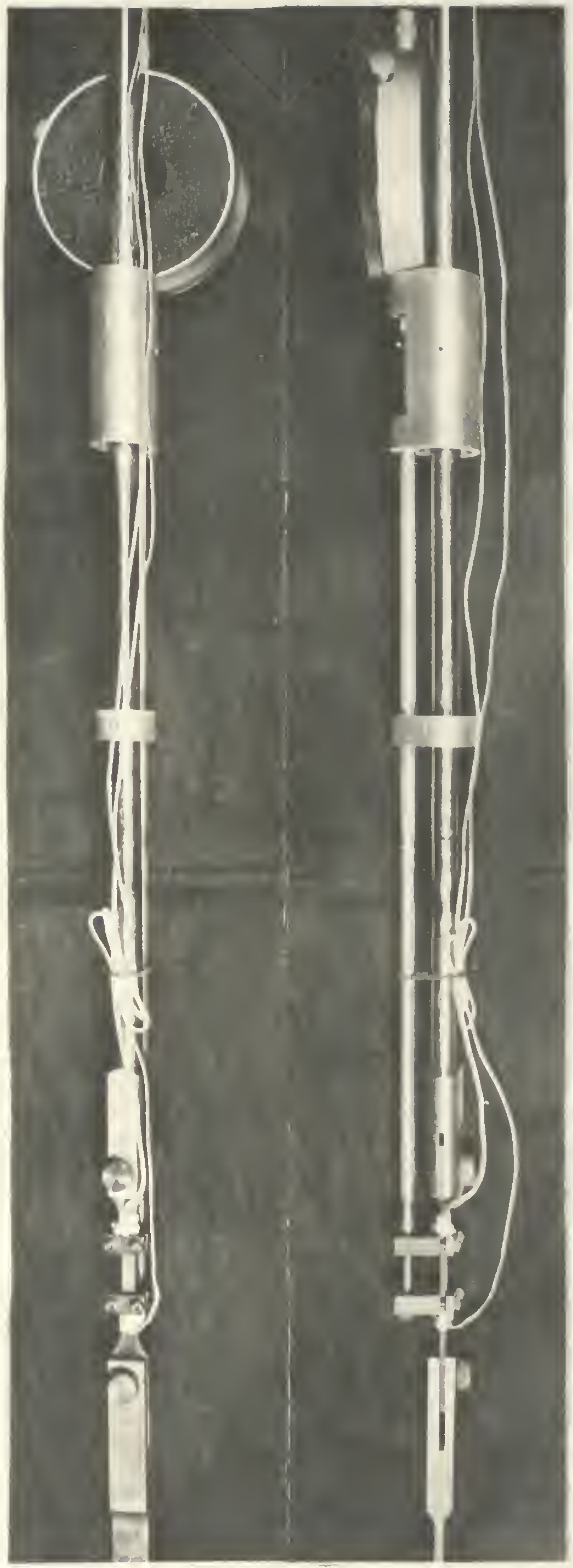


Fig. 7 **Extensometer Unit**

winding permitted zone-by-zone heat adjustment. This regulation was accomplished by attaching independently variable external shunts to these taps thereby permitting the by-passing of a portion of the current in any desired section of the furnace. Indicating temperature controllers, each actuated by a chromel-alumel thermocouple, controlled the on-and-off power input. To minimize the amplitude of temperature cycling, a resistance was placed in parallel with the control relay. This enabled the current supplied to the furnace to be varied between maximum and minimum values differing by 20%. The 20% value was necessary to overcome changes in ambient temperature and variation in packing around the pulling tab from test to test. Voltage regulators served to keep the voltage variation to less than 0.1%. In this manner the temperature was controlled to within $\pm$ 2°C. with not more than 1°C. variation along the gage-length.

3. Experimental Technique

The constant-load creep tests were carried out on specimens 1/10 in. thick by 5/32 in. wide which had a reduced section of 1-3/8 in., the central 1-inch of which constituted the gage section. The specimen was attached to the extensometer system by two 1/4-inch diameter pins. Calibrated thermocouples were loosely bound at each end of the gage length by fiberglass string. The extensometer and specimen assembly was then lowered into the 2½-inch internal diameter furnace tube, which was slightly above the test temperature. The furnace ends were closed with asbestos cloth to prevent free circulation of air that would produce undesirable temperature variations. The test temperature was attained in approximately one hour, the load was then applied, and readings were started. Readings were taken frequently during the first 2 to 3 hours, and afterwards at increasing intervals. A microswitch, which was connected to the frame of the creep unit, permitted obtaining time-to-rupture data. The contact in the switch was opened by the lever arm falling into the "unloaded" position causing the timing circuit to be interrupted.

Total extension (elastic plus plastic strain) was plotted against time, and minimum creep rates were calculated by determining the slope of the straight portion of the resulting curves.

A section of the fracture, of a representative number of specimens, was mounted in thermo-setting Bakelite for metallographic examination, together with a strip of material from a

less-stressed area of the grip section which had been exposed at the test temperature. Samples of the five initial states before testing were also prepared for metallographic examination. All samples were etched with a potassium bichromate reagent.

4. Specimen Design Problems

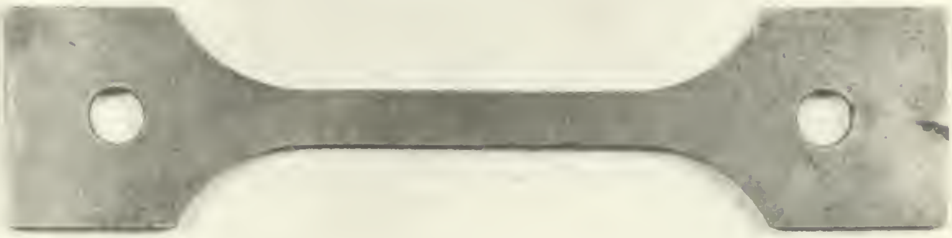
At the beginning of this investigation it was assumed that the customary creep specimen design (see Figure 8c), being employed in the Creep Testing Laboratory of the U. S. Naval Postgraduate School, would be adequate. Upon completion of the first test on a specimen, in the HT-316 condition, based on this design it appeared that this assumption was justified as the fracture occurred within the gage section. However, the next four fractures occurred at the grip section as shown in Figure 9c). These tests were conducted on specimens in the H and HT-316 conditions.

It was recognized that these inconveniently located fractures were due to the stress concentration caused by the presence of the centrally located circular hole, and further aggravated by the brittleness of the material, in that local yielding was not possible.

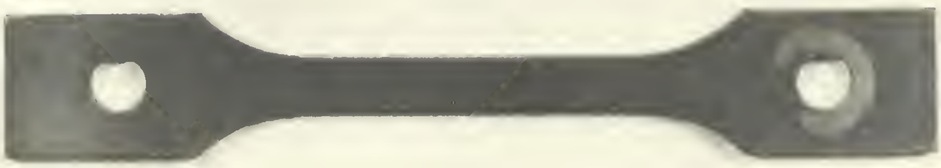
This, obviously, meant that a design change was in order. A 60% increase in the width of the grip section was considered sufficient; this modification constituted the intermediate design as shown in Figure 8b. Four tests were then conducted using specimens of such design, in the A, H and HT-316 conditions. All of these tests resulted in fractures in the same location (see Figure 9b) as in the previous four tests.

At this stage of the investigation, the writer's thoughts were, "Would these designs prove adequate when subjected to short-time tensile tests at room temperature?". The results of such tests are shown in Figure 10. Of course, the next logical

(a)



(b)



(c)

Figure 8 - Test Specimens

(a) Final Design
(b) Intermediate Design
(c) Initial Design

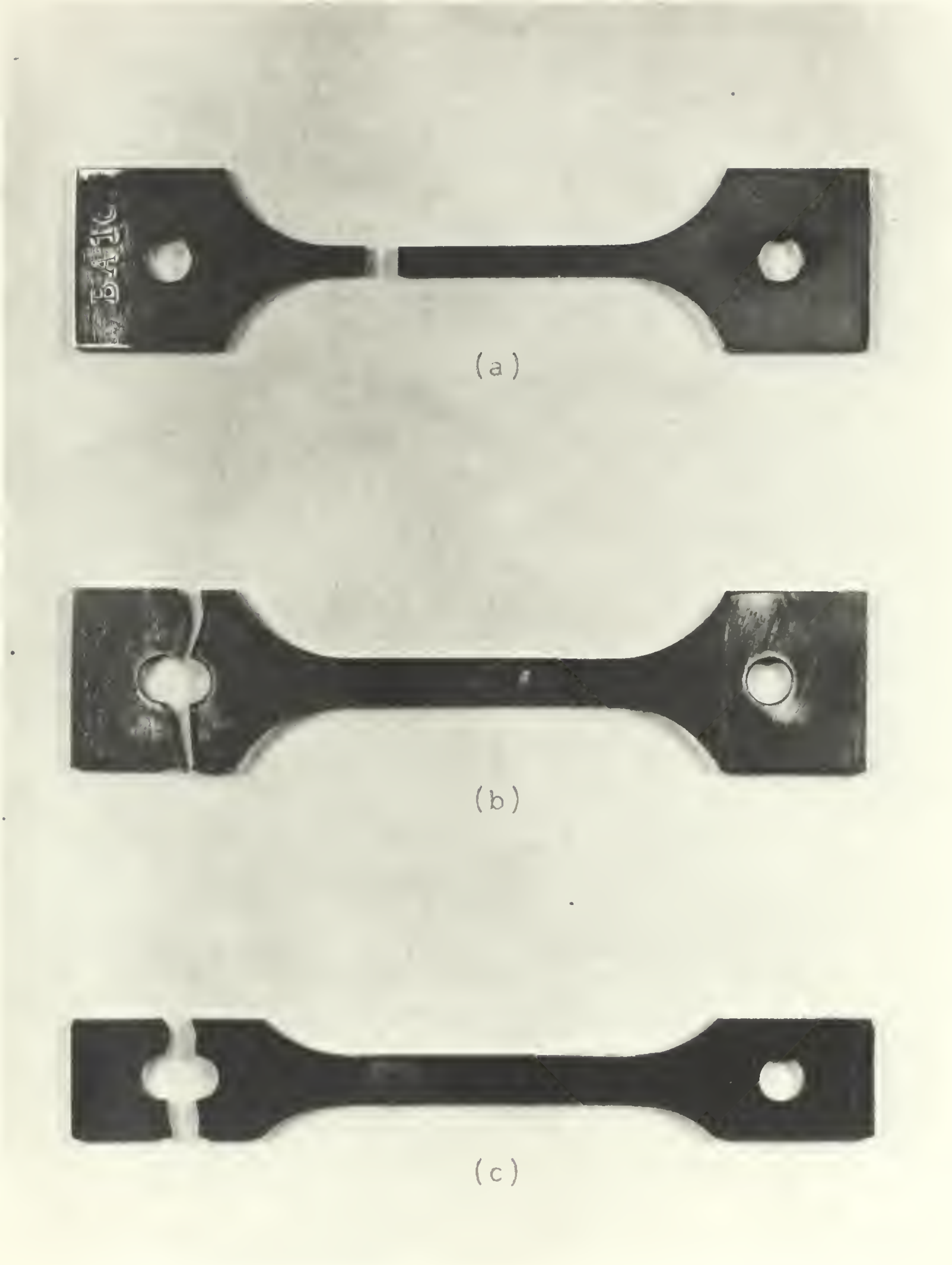
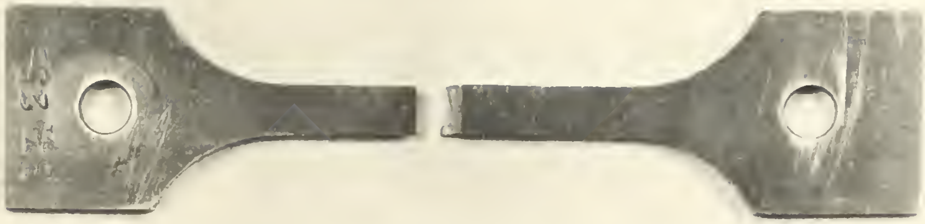


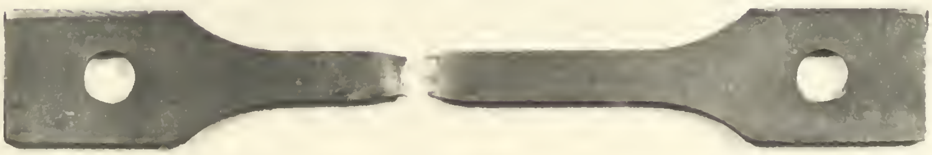
Figure 9 - Test Specimens Showing Location of Fractures Under Creep Tests at 300°C.

(a) Final Design, AT-316 Condition
(b) Intermediate Design, HT-316 Condition
(c) Initial Design, AT-316 Condition



(a)

(b)



(c)

Figure 10 - Test Specimens Showing Location of Fractures Under Short-Time Tensile Tests at Room Temperature

(a) Intermediate Design, HT-316 Condition
(b) Intermediate Design, H Condition
(c) Initial Design, H Condition

step in this development would be to conduct the short-time tensile tests at 300°C. However, due to time considerations, it was felt that another modification would resolve the situation. Accordingly, the width of the gage section was reduced by 37.5% resulting in the final design as shown in Figure 8a. Pertinent dimensions of all three test specimens are shown in Table III.

TABLE III - DIMENSIONS OF TEST SPECIMENS

Design	Grip Width, in.	Reduced Section Width, in.
Initial	5/8	1/4
Intermediate	1.0	1/4
Final	1.0	5/32

While this final change did cause the location of the fractures to shift to the reduced section (see Figure 9a), it did not solve the problem entirely. A summary of the location of fractures utilizing this design is shown in Table IV.

TABLE IV

SUMMARY OF FRACTURE LOCATIONS UTILIZING FINAL DESIGN SPECIMENS IN CREEP TESTS AT 300°C.

Location	Number
In Gage Length	8
At Edge of Gage Block	16
At Gage Point	12
Outside Gage Length	1

Due to the predominance of fractures at the gage blocks and gage points, there arose a suspicion as to whether or not the creep data obtained from such tests were valid. In order to justify the use of such data, eight creep-rupture tests (i.e., specimens were prepared in a manner similar to that described above for the creep tests, except that the strain measuring system was omitted), were conducted. All of these specimens fractured in the same location normally occupied by the gage blocks. It was therefore concluded that the location of failure was not due to gaging technique and that the creep data accumulated could be considered fairly reliable.

If time permitted further work to be undertaken, the next modification might have been to have the specimens made with a gradual taper from both ends of the reduced section to its mid-length as mentioned by Davis, Troxell, and Wiskocil[8].

EXPERIMENTAL RESULTS

Table V summarizes the results of the creep tests conducted during this investigation. Hardness measurements are included for the purpose of possibly evaluating the rate, if any, at which precipitation proceeded and to determine if this rate was accelerated by strains introduced during testing.

Figure 11 shows a plot of the stress versus the log of the minimum creep rate, while Figure 12 shows a plot of the stress versus the log of the rupture time. The literature shows that such data are commonly plotted either on a double log basis or a semilog basis[9, 10]. The data of Parker and Ferguson[5] are included in Figure 12 (dashed lines) for comparison. The above authors conducted rupture tests at 350°C. on beryllium-copper (2.25% Be.) wires in the half-hard temper and pre-aged at 300°C. for 1½ hours.

TABLE V - RESULTS OF CREEP TESTS AT 300°C. (572°F.)

Initial Stress, psi	Condition	Initial Extension, %	Minimum Creep Rate, %/hr.	Rupture Time, hrs.	Total Strain at Fracture, %	Rockwell Hardness		
						Before Test	After Test Grip Section	After Test Reduced Section
45,000	A	0.14	0.021	61.12	1.89	B 76	C 39.5	C 38.5
	AT-316	0.23	0.014	98.0	2.34	C 40.5	C 38.5	C 37.5
	AT-600	2.71	0.080	62.5	12.5	B 70.5	B 74.5	B 90
49,000	A	0.13	0.030	36.25	1.14	B 76	C 41.5	C 40.5
	AT-600	4.10	0.36	18.13	15.63	B 70.5	B 72.5	B 89
	H	0.18	0.006	460.0	5.22	C 21.5	C 40	C 35.5
	HT-316	0.04	0.007	452.5	4.89	C 45	C 40	C 35.5
53,000	A	0.31	0.065	12.6	1.62	B 76	C 41.5	C 40.5
	AT-316	----	0.025	16.65	1.56	C 40.5	C 41	C 40.5
	AT-600	5.80	2.0	7.2	18.75	B 70.5	B 71.5	B 89
57,000	AT-316	0.32	0.048	13.25	1.27	C 40.5	C 40.5	C 40
	AT-600	6.20	6.8	1.23	18.75	B 70.5	B 70.5	B 89.5
60,000	A	0.32	----	6.7	1.56	B 76	C 40.5	C 40.5
	A	0.43	----	4.7	1.18	B 76	C 40	C 34
	A	0.49	0.12	3.17	1.26	B 76	C 40.5	C 38.5
	A	0.28	----	5.88	1.56	B 76	C 40.5	C 41.5
	AT-316	#	#	11.64	#	C 40.5	C 41.5	C 40.5
	AT-600	12.1	25.1	0.38	25.0	B 70.5	B 70.5	B 90.5
	H	0.39	0.016	142.3	4.85	C 21.5	C 42	C 38
	HT-316	0.30	0.018	128.8	4.52	C 45	C 42	C 39.5
65,000	A	0.33	0.17	2.48	0.86	B 76	C 39	C 39
	A	0.22	0.19	2.81	1.05	B 76	C 41	C 40
	AT-316	#	#	5.65	#	C 40.5	C 41	C 41.5
70,000	A	0.35	0.29	0.74	0.62	B 76	C 37.5	C 38
	AT-316	#	#	1.30	#	C 40.5	C 41.5	C 40.5
	H	0.35	0.046	35.45	3.2	C 21.5	C 44	C 43
	HT-316	0.43	0.058	31.9	2.77	C 45	C 44.5	C 42
75,000	A	0.36	0.50	0.38	0.52	B 76	C 37	C 38
80,000	AT-316	#	#	0.11	#	C 40.5	C 41	C 41
	AT-316	#	#	0.14	#	C 40.5	C 41	C 41
	H	0.49	0.12	12.25	2.9	C 21.5	C 44.5	C 43
	HT-316	0.52	0.14	16.50	2.8	C 45	C 44	C 42.5
	HT-316	#	#	16.75	#	C 45	C 45	C 44
	HT-316	#	#	11.85	#	C 45	C 45	C 43.5
90,000	H	0.36	0.44	3.08	1.93	C 21.5	C 44.5	C 44
	H	#	#	3.02	#	C 21.5	C 44	C 43.5
	HT-316	0.54	0.25	2.72	1.53	C 45	C 44.5	C 43.5
	HT-316	0.41	0.25	3.80	1.71	C 45	C 44	C 43

# Denotes creep-rupture tests

[illegible]	A	0.35	0.29	0.74	0.62	R 76	C 37.5	C 38
	AT-316	#	#	1.30	#	C 40.5	C 41.5	C 40.5
	H	0.35	0.046	35.45	3.2	C 21.5	C 44	C 43
	HT-316	0.43	0.058	31.9	2.77	C 45	C 44.5	C 42
75,000	A	0.36	0.50	0.38	0.62	R 76	C 37	C 38
80,000	AT-316	#	#	0.11	#	C 40.5	C 41	C 41
	AT-316	#	#	0.14	#	C 40.5	C 41	C 41
	H	0.49	0.12	12.25	2.9	C 21.5	C 44.5	C 43
	HT-316	0.52	0.14	16.50	2.8	C 45	C 44	C 42.5
	HT-316	#	#	16.75	#	C 45	C 45	C 44
	HT-316	#	#	11.85	#	C 45	C 45	C 43.5
90,000	H	0.36	0.44	3.08	1.93	C 21.5	C 44.5	C 44
	H	#	#	3.02	#	C 21.5	C 44	C 43.5
	HT-316	0.54	0.25	2.72	1.53	C 45	C 44.5	C 43.5
	HT-316	0.41	0.25	3.80	1.71	C 45	C 44	C 43

# Denotes creep-rupture tests

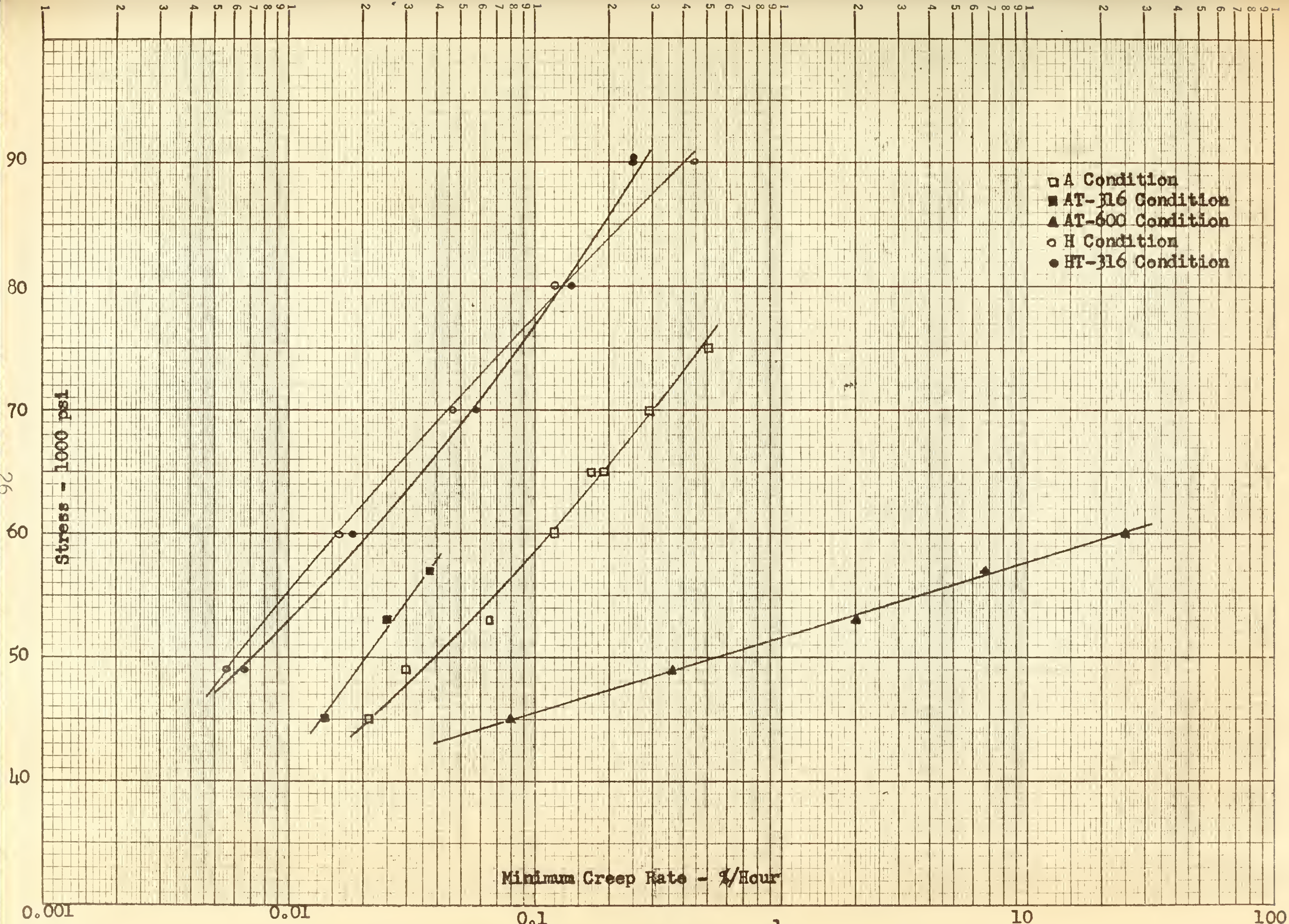


Figure 11 - Semilog Plot of Stress versus Minimum Creep Rate

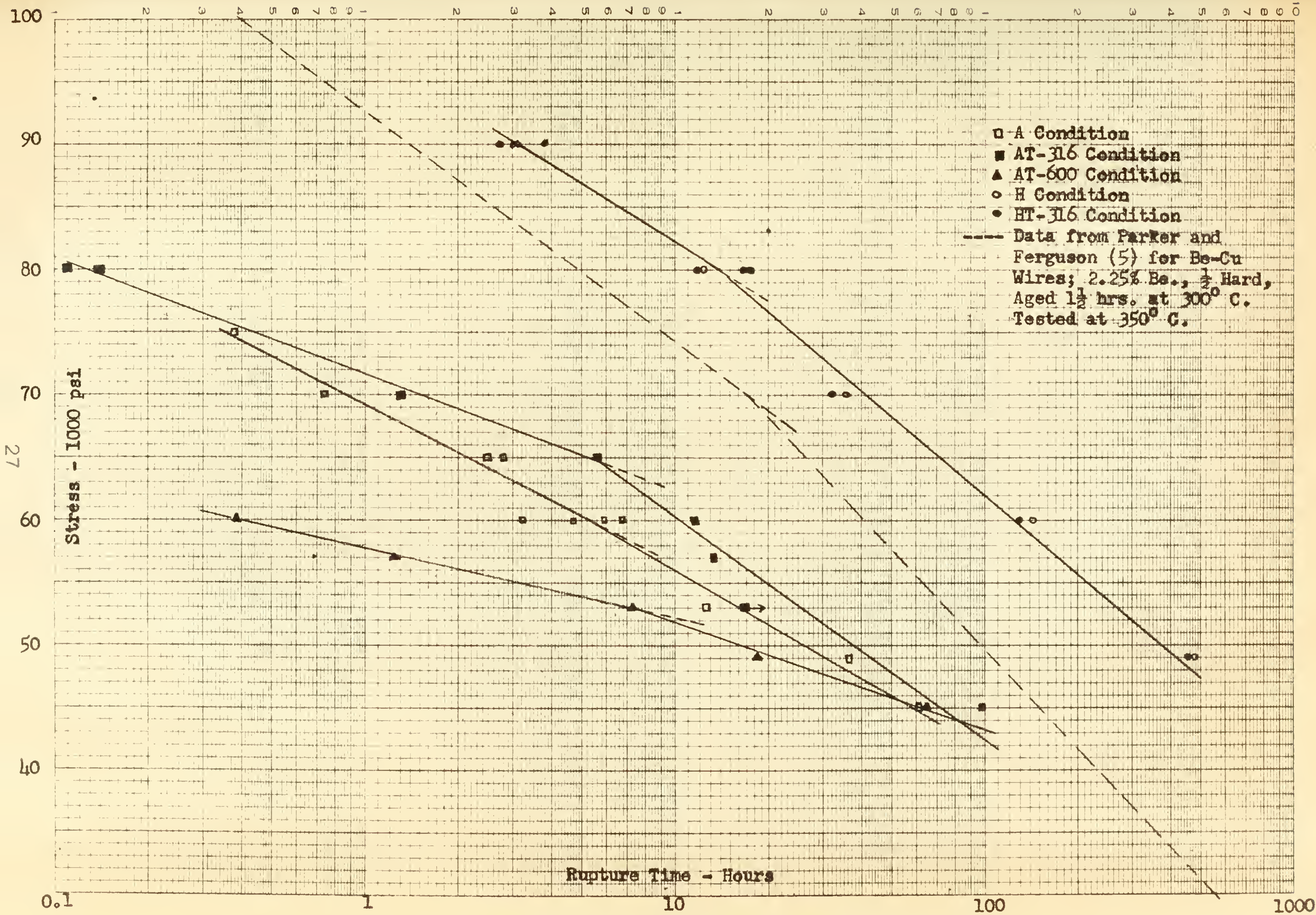


Figure 12 - Semilog Plot of Stress versus Rupture Time

DISCUSSION OF RESULTS

It has been shown[11, 12] that the precipitation process may involve several reactions:

(1) localized precipitation at grain boundaries and slip lines.

(2) general precipitation within the grains and coalescence at boundaries.

(3) discontinuous or heterogeneous precipitation.

(4) coalescence and spheroidization of the second phase.

It is not the purpose here to discuss these reactions, but to point out briefly the meanings of the above terminology before proceeding with its use in this discussion.

As would be expected, the first precipitate to be detected is localized in regions of relatively high free energy such as grain boundaries and slip bands. The appearance of this localized type of precipitation as occurs at grain boundaries is shown in Figure 3b. General (continuous) precipitation implies merely a uniform distribution of particles within the general grain areas (see Figure 1a). The type of precipitation that is characterized by the formation of more or less lamellar nodules of the two-phase mixture of matrix plus precipitate (see Figure 2b) has been referred to as discontinuous or heterogeneous precipitation. The precipitation process may be considered complete when the matrix and second phase have assumed their equilibrium compositions and crystal structures. This involves

coalescence and spheroidization of the second phase as shown in Figure 5b. Geisler[11] points out that many of these precipitation reactions may be occurring in the alloy at different areas simultaneously, or they may start after widely different incubation periods.

Metallographic examination of specimens in the A and AT-316 conditions revealed that the discontinuous type of precipitate at the grain boundaries had occurred here, the volume of this precipitate being greater in the aged (AT-316) condition. A comparison of these specimens with fractured specimens showed that the discontinuous precipitate increased in volume (see Figure 13), growing into the grains, during the creep test. However, no significant difference in the amount or size of the general precipitate within the grains was observed between the four conditions, A, AT-316, A-ruptured and AT-316-ruptured. In connection with the ruptured specimens it was noted that the volume of the discontinuous precipitate was greater in specimens having longer life-to-rupture (lower stresses). This was believed to be due to essentially time at temperature as no significant differences were observed between areas in the reduced section and lightly stressed grip section. Also, the initial difference in appearance between A and AT-316 conditions was probably continually lessened with time under stress at 300°C. as suggested by the similarity between A-ruptured and AT-316-ruptured specimens.

In the H and HT-316 conditions localized precipitation at the grain boundaries was always observed, the coarseness in-

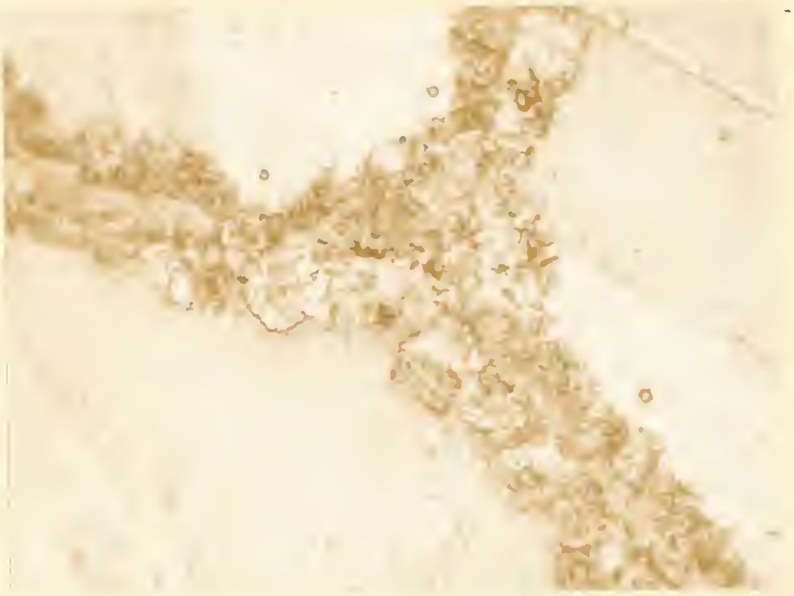


Fig. 13--Photomicrograph of the A Condition Shown in Fig. 1 after being Tested to Fracture at 45,000 psi at 300°C. Rupture Time: 61 hrs. X2000.

Fig. 14--Photomicrograph of the H Condition Shown in Fig. 3 after being Tested to Fracture at 49,000 psi at 300°C. Rupture Time: 460 hrs. NOTE GRAIN BOUNDARY CRACK. X2000

creased (see Figure 14) with increasing creep-rupture life. This type of precipitation was associated with the optimum creep resistance in the stress range covered by this investigation as shown in Figure 12. There is no marked difference in the creep characteristics of the H and HT-316 conditions as shown in Figures 11 and 12. The H condition appears to have a very slight superiority over the HT-316 condition. Whereas in the A and AT-316 conditions the trend is reversed and the difference between these two conditions is greater but still not appreciable.

The nature of the fractures and the mode of deformation of the specimens in the A, AT-316, H and HT-316 conditions were of the high temperature type. That is, the fractures were all intergranular and deformation was devoid of visible distortions within the grains. At high stresses the H and HT-316 conditions showed considerable intergranular cracking approximately normal to the direction of stressing over the entire gage length, while at low stresses this cracking was restricted to the immediate zone of the fracture.

Geisler[11] states that the loss in ductility and the development of an intergranular fracture in precipitation alloys have been associated with the presence of a precipitate at grain boundaries. This also applies to fracture through the grain-boundary nodules of recrystallized matrix, as in the A and AT-316 conditions, which exhibit the grain boundary reaction early in the aging sequence while the grains are still hardening.

Material in the AT-600 (over-aged and slowly cooled) condition, which was structurally stable at test temperature, showed the occurrence of mixed fractures during a period of transition from entirely transgranular at high stresses to entirely intergranular with some internal cracking at low stresses. It is evident that the short-time creep strength of material in this condition is poorer than material in any other condition. However, if the trends indicated by the stress-rupture curves of Figure 12 continue, it would appear that upon extrapolation of these curves for longer rupture times (lower stresses), the A, AT-316, H and HT-316 conditions will lose their superiority to the AT-600 condition. This is in agreement with Jenkins et al.[2] and Dennison[4] who showed that the slowly cooled condition was superior to the pre-aged or solution-treated state at low stresses.

Metallographic examination and hardness measurements of the fractured specimens did not yield any conclusive evidence that strain-induced aging took place. A specimen which had been aged, in an unstressed condition, at 300°C. for 61 hours showed the same type and amount of precipitation as a specimen that had been creep tested to fracture at the same temperature and a rupture life of the same duration. The marked increase in hardness in the reduced section (see Table V) of the AT-600 specimens is undoubtedly the result of strain hardening.

The data obtained in this investigation appears consistent with that of Parker and Ferguson[5] when due consideration is given to the initial conditions and test temperatures.

CONCLUSIONS

From the results obtained in this investigation, the following conclusions can be drawn:

1. In short-time creep tests, the cold worked material shows its superiority over material in any other condition.

2. The effects of pre-aging were not marked in improving the creep properties of beryllium-copper when tested at 300°C.

3. The data suggests that for long-time tests the fully softened (AT-600) condition should be superior in strength over material in any other condition.

4. Because of the marked difference in ductility in specimens after various heat treatments, comparisons of the minimum creep rates do not give a good indication so far as life to rupture is concerned. Material in the fully softened (AT-600) condition shows fairly good ductility, whereas little elongation is found in other conditions. When, for example, these materials are loaded at stresses which result in an identical rate of creep, the life to rupture of the fully softened material is considerably greater than that of material treated in other ways.

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